

XVII International Colloquium on Mechanical Fatigue of Metals (ICMFM17)

Experimental and numerical analysis of short fatigue cracks in Lean Duplex Stainless Steels

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Abstract

Transmission (TEM) and scanning electron microscopy (SEM) in combination with automated electron back-scattered diffraction (EBSD) techniques is used to study the surface damage as well as the initiation and propagation of short fatigue cracks in LDSS AL 2003 in two different thermal conditions. Whereas in the as received (AR) condition the initial deformation lines appear in the ferrite phase, in the heat treated (HT) condition the deformation lines appear in both phases at the same time. On the other hand, in both thermal conditions short cracks nucleate along intrusion/extrusions in the ferrite phase. Additionally, applying a numerical model it was found that the path of short cracks not only depends on the Schmid Factor but it is also conditioned by the angles of the crack-plane deflection at a grain boundary.

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Selection and peer-review under responsibility of the Politecnico di Milano, Dipartimento di Meccanica

Keywords: low cycle fatigue; lean duplex stainless steel; short crack propagation

1. Introduction

Duplex stainless steels (DSS) have a ferrite matrix (α) intermixed with austenite (γ). The duplex microstructure combines the high strength of ferrite with the good ductility and corrosion resistance of austenite. Nowadays, DSS new developments are focused on reducing costs, giving rise to a low-alloyed grade named Lean DSS.

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Nomenclature

DSS	duplex stainless steel
LDSS	lean duplex stainless steel
LCF	low cycle fatigue
HCF	high cycle fatigue
AR	as received
HT	heat treatment
EBS	electron backscatter diffraction
SF	Schmid factor
TEM	transmission electron microscopy

The microstructural features of DSS, such as chemical composition, volume fraction of phases, phase distribution, grain size and heat treatment, have a great influence in its properties [1]. In particular, in standard DSS [2, 3] and LDSS [4, 6] microcrack nucleation during fatigue depends on the near-surface microstructure and its crystallography while microcrack propagation is affected by microstructural barriers. The purpose of this work is to analyze in LDSS AL 2003 the influence of microstructural changes, produced by thermal treatments, in the cyclic plastic slip activation, short crack nucleation and propagation.

2. Material and experimental procedure

The investigated material was the LDSS AL 2003 (UNS S32003). The chemical composition of the present LDSS is: C:0.021; Cr:22; Ni:3.8; Mo:1.8; Mn:1.73; Si:0.22; P:0.024; N:0.18; Fe: balance in weight percent. This steel was received in longitudinally welded stainless steel pipe. Specimens which received no further heat treatment are named as received (AR). A heat treatment (2h at 1250°C) followed by a solution treatment at 1050°C and subsequent water quenching was carried out to increase the grain size of the AR steel [6]. The material with coarse grains are referred as heat treat (HT) material. Table 1 shows the grain size, determined by EBSD technique, and Vickers hardness of each phase in both thermal conditions. From slabs taken parallel to the axis of the pipe, flat specimens with notch for low cycle fatigue (LCF) were machined with 20 mm gauge length and 25 mm² area. All the specimens were initially ground and polished with sequentially finer grits paper and finally were electrolytic polished. Cyclic tests were conducted at room temperature under fully reversed plastic strain control, with a plastic strain range of $\Delta\epsilon_p = 0.2\%$. In order to observe the damage evolution during LCF tests, a central sector of the specimens was selected. Images of this zone were acquired before the test. Then, cycling was systematically interrupted and the selected area was recorded. The observation during tests was done using the optical system detailed in [4].

Table 1. Grain size and Vickers hardness of LDSS.

Material	Phase	Grain size (μm)	Vickers hardness
AL2003AR	Ferrite	7.0 ± 4	310 ± 10
	Austenite	4.4 ± 2	320 ± 20
AL2003TT	Ferrite	10 ± 19	273 ± 20
	Austenite	20 ± 16	273 ± 15

3. Results and discussion

A detailed study of fatigue surface damage initiation shows differences between AR and HT specimens. Whereas in the AR condition the first deformation lines appear in the ferrite phase, in the HT condition slip markings arise simultaneously in both phases. In both conditions, AR and HT, as cycling proceeds short cracks nucleate on extrusions/ intrusions in ferritic grains. However, the origin of the short crack nucleation differs in each case. In AR samples, Strubbia et al [5] reported that the microcracks nucleation in ferrite can be rationalized by a larger plastic

activity found in this phase in comparison with austenite. In the present work, this cyclic plasticity partition is ascribed to the bigger grain size and less hardness of ferrite than austenite, Table 1. In HT samples, the austenite phase sustains most of the plastic deformation, Fig. 1, as a consequence of its bigger grain size than ferrite, Table 1. However, microcracks nucleation occurs in the ferrite phase. This fact can be attributed to Cr_2N precipitation in the ferrite phase. In DSS and superferritic stainless steels [7,8], thermal treatments at high temperature followed by water cooling produce needle-like Cr_2N precipitates that detrimentally affect mechanical properties. This precipitation reduces dislocation motion and degrades ductility. Therefore, it can be presumed in this work that in LDSS AL2003 HT the precipitation of Cr_2N embrittles the ferrite phase leading to earlier cracks nucleation.

As regards crack propagation, in both AR and HT specimens, Strubbia et al [6] observed two alternative crack growing mechanism when a crack nucleated in the ferrite phase reaches an austenite grain: i) along a favorable slip plane with high SF (stage I) or ii) alternating between two slip systems (stage II). These authors also calculated the plane of the crack in stage II and found that some cracks grow by double slip on lattice planes of $\{1\ 1\ 0\}$ type. Additionally, they reported that independently of the growing mechanism (stage I or II), cracks grow approximately perpendicular to the specimen axis.

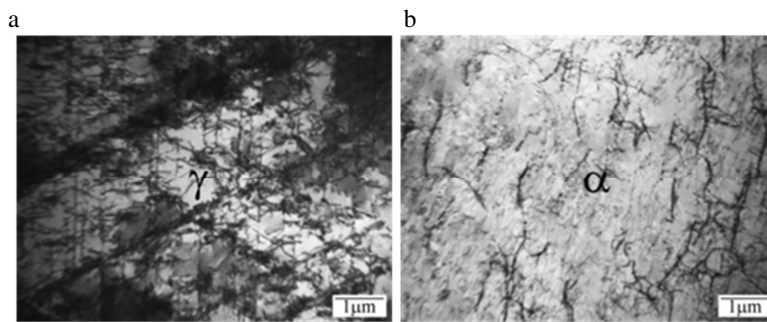


Fig. 1. Dislocation structure in AL 2003 HT during LCF, 5000 cycles, (a) austenite and (b) ferrite.

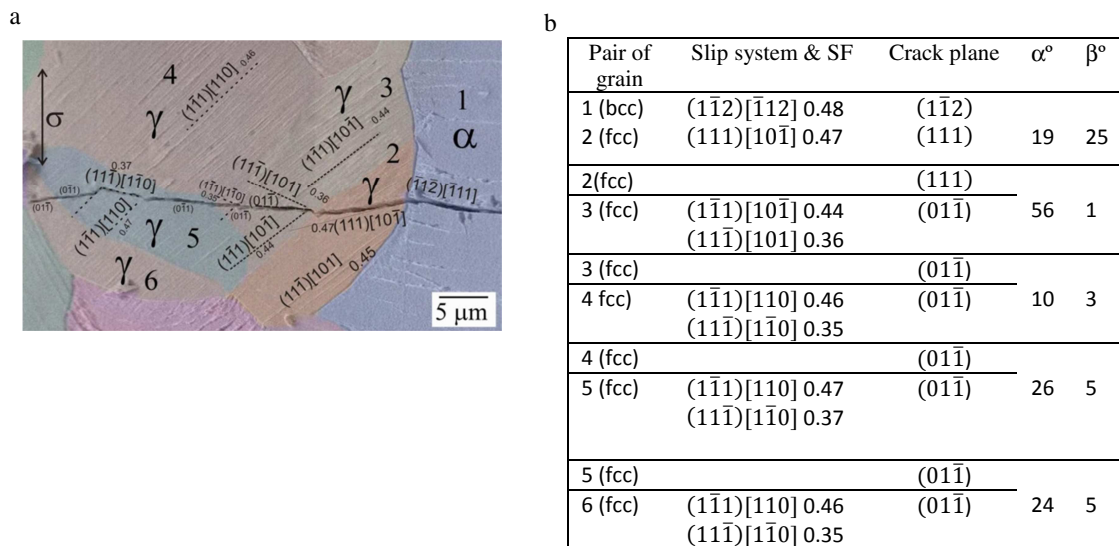


Fig. 2. (a) EBSD Euler map, slip systems and crack plane in AL2003 HT (b) Analysis of slip systems with SF, twist (α) and tilt (β) angles of the crack-plane deflection at a grain boundary.

It is known that the evolution of surface roughness, subsequent nucleation and growth of short fatigue crack in DSS is strongly influenced by the microstructural parameters, i.e., grain size, grain orientation, grain and phase boundary geometry and precipitates [9]. Consequently, the growth rate of microstructurally short cracks is determined by the crack-tip slide displacement and depends on the resistance against dislocation motion on the slip band ahead of the crack and across microstructural barriers. The main parameters that describe the resistance are the length of the slip band and the local crystallographic misorientation conditions between the slip bands of neighbouring grains. According to Zhai et al. [10], the twist (α) and tilt (β) angles of the crack-plane deflection at a grain boundary are the key factors that control the path and growth rate of a short crack. The smaller the twist angle α and the tilt angle β are, the lower is the resistance to microcrack propagation. The angles α and β were calculated for each slip system and crack plane exhibiting similar behaviour in AR and HT specimens. Fig. 2a shows in the ferrite phase the microcrack grows in stage I but when it arrives to the neighbouring austenitic grain follows a single slip system with high SF, in this case 0.47 (Fig.2b). Thereafter, the microcrack propagates along austenitic grains in planes with a small relation of α and β , which are not necessary slip planes (Fig. 2b). This last result justified the occurrence of crack propagation by stage II.

4. Conclusion

In this work the microstructure, cyclic plastic slip activation, short crack nucleation and propagation were studied in LDSS AL 2003 in different thermal conditions, AR and HT. The results have revealed that in the AR condition the first deformation lines appear in the ferrite phase and in the HT condition slip markings arise simultaneously in both phases. Moreover, with respect to crack initiation site in both conditions, AR and HT, short cracks nucleate on extrusions/ intrusions in ferritic grains. This microcracks nucleation in ferrite, in AR specimens is due to a larger plastic deformation developed in this phase whereas in HT condition can be attributed to Cr_2N precipitation observed in ferrite. On the other hand, the crack propagation in both AR and HT specimens corresponds along a favorable slip plane with high SF (stage I) or alternating between two slip systems (stage II). The occurrence of crack propagation by stage II can be explained by the small relation of α and β .

Acknowledgements

This work was supported by Consejo Nacional de Investigaciones Científicas y Técnicas (CONICET).

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